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Chromium Steel Castings

FOR many years there has been severe competition between steel castings and welded or forged materials of construction. Yet at present in many fields castings are more than holding their own. This is due to several factors which have contributed to the development of steel castings, as well to certain advantages which cast steel possesses over welded construction in components subject to severe conditions of service, such as vibration or repeated alternations of temperature. Modern development of steel castings has been due to intensive study of foundry technology, to the increased use of electric furnaces, to the improved methods of heat treatment now applied, and to the use of alloy steel castings to meet the highest requirements and sometimes to replace the more expensive castings of non-ferrous metals.

These factors are illustrated in a paper by Dr. K. Roesch[1] on "Chromium Steel Castings," in which he shows what a wide range of properties can be imparted to the steel by the addition of different percentages of chromium. The addition of 1 or 2 per cent. of chromium increases the strength sufficiently to allow a saving of weight, while response to heat treatment is further improved by the addition of molybdenum. With 6 per cent. and upwards of chromium the steel becomes heat-resisting and with 12 to 16 per cent. and a low carbon content it is corrosion-resistant. A cast steel of this type is superior to a bronze casting, especially when high strength and wear resistance is required as well as resistance to corrosion.

Commercial chromium steel castings are grouped by the author according to Table I :—

TABLE I.—*Composition of Chromium Steel Castings*

Group.	C, per cent.	Cr, per cent.	Mo, per cent.	Use.
I*a*	0·12 to 0·20	0·80 to 1·50	0·25	Case-hardening
I*b*	0·20 to 0·35	0·80 to 2·00	0·25	For heat treatment
II*a*	0·15 to 0·25	About 1·0	About 0·5	High pressure fittings
II*b*	0·15 to 0·25	3·0 to 6·0	About 0·5	and in the oil industry
III	0·15 to 0·60	6–16	—	Heat resistance
IV	0·15 to 0·30	14–16	—	Corrosion resistance

Group I*a* consists of low-carbon constructional steels, which incidentally have good magnetic properties, but are most suitable for cog wheels, sprocket wheels, &c., which are to be case-hardened. Group I*b* are stated by Roesch to be used for cast steel parts in aircraft and conform to the German Aircraft Specification 1811, their good properties actually leading to a considerable saving in weight. The steel is also weldable and in this respect the acid steel is said to be better than the basic. For an average composition the mechanical properties, given in the form of curves, are as recorded in Table II. The advantages

TABLE II.—*Mechanical Properties After Heat Treatment*

	Average Type I*b*.			Type IV, 14 per cent. Cr.		
Tempering temperature, deg. C.	500	600	700	500	600	700
Yield point, tons per square inch	63	50	39	67	50	41
Maximum load, tons per square inch	76	62	51	76	57	51
Elongation per cent ($l = 5\,d$) ...	9	15	21	8	10	14
Reduction of area, per cent. ...	30	42	55	—	20	40
Impact figure, mkg./cm.^{2}	—	10	15	—	2	6

of the addition of molybdenum in promoting depth hardening and in allowing the use of higher tempering temperatures to obtain a given tensile strength with increased ductility are emphasised. The steels of group II*a* are used in high-pressure steam plant at temperatures up to 500 deg. Cent. They are notable for their creep resistance, which is enhanced by the presence of molybdenum. The creep limits at 500 deg. of carbon steel, a steel with 0·5 per cent. of molybdenum, and one with 1 per cent. of chromium and 0·5 per cent. of molybdenum are given as 7 kilos., 12 kilos., and 16 kilos. per square millimetre, respectively, *i.e.*, 4·4, 7·6, and 10·1 tons per square inch. A casting of a high degree of heat and corrosion resistance is that of type II*b*, largely used in oil refinery work at temperatures up to 600 deg. Cent. It is also immune from hydrogen embrittlement, since the carbide consists mainly of chromium carbide, which is not attacked by hydrogen. Group III contains the heat-resisting steels, frequently used with higher chromium and rather higher carbon content than that given. The maximum admissible temperature at which they may be exposed to furnace gases is stated to rise 18 deg. for every 1 per cent. of chromium, from about 700 deg. Cent. for 6 per cent. of chromium to about 1130 deg. Cent. for 30 per cent. of chromium. Group IV contains the corrosion-resisting castings in which it is essential that the chromium should be over 11 per cent. (preferably 14 per cent.) and the carbon low. A hardening temperature of 950 deg. to 1050 deg. Cent. must be used. The properties of the 14 per cent. chromium steel castings of this type are given in the form of curves from which the results recorded in Table II have been derived.

The author is justified in taking chromium as the element in alloy steel castings which can confer the widest range of properties. For some uses it could no doubt be replaced by other, and usually more expensive, alloy elements, but without chromium the whole field of the most valuable heat-resisting and corrosion-resisting steels would be non-existent.

REFERENCE

[1] *Die Giesserei*, July 14th, 1939, Vol. XXVI, page 357.

Columnar Structure in Steel Ingots

UNDERLYING nearly all discussions of the process of solidification is the classical theory of Tammann according to which crystallisation begins in a pure liquid by the formation of isolated crystal nuclei, and the final form and number of the resulting crystallites is dependent on the number of nuclei and the velocity of crystallisation. There are differences of opinion as to the manner in which the principle of Tammann may be applied to the explanation of the solidification of a steel ingot. Any picture of the process in terms of nucleus number and crystallisation velocity, and the mutual relations between them, is complicated by the necessity of allowing for the influences of casting temperature, rate of casting, heat flow in the mould, and chemical composition of the steel which governs micro-segregation. From the conflicting views expressed by different authors there appear to be special difficulties in the way of a satisfactory explanation of the formation of a columnar structure in a steel ingot. Some regard the persistence of columnar growth as due to undercooling, since pronounced undercooling is associated with a low nucleus number and a high velocity of crystallisation. Others come to the conclusion that undercooling cannot be present in the region of columnar crystal growth. The ingot is bordered by a very narrow fringe of fine-grained structure, and it is held that undercooling is, on the whole, improbable after the solid phase has once formed and would not be likely to persist long on account of the liberation of latent heat of solidification. Columnar crystals, springing from the first formed layer of fine-grained material, continue to grow so long as there is a favourable temperature gradient, but when the temperature fall in the liquid adjacent to the growing crystal is arrested by emission of the latent heat and by conduction from the hot interior, the velocity of growth of the columnar crystal is reduced and crystallisation takes place at numerous centres in the remaining liquid giving rise to an equiaxial structure. Susceptibility to columnar growth is enhanced by maintaining a steep temperature gradient in the cooling metal, *i.e.*, by arranging for a rapid heat flow from metal to mould, and it also varies with composition of the steel. Thus, E. Houdremont found that columnar growth appeared to be strongest in the austenitic nickel and nickel-chromium steels which solidify immediately into the γ region and in the high-carbon martensitic-pearlitic nickel-chromium steels.

In a paper entitled "A Contribution to the Problem of the Solidification of Steel Ingots," R. Hohage and R. Schäfer[1] call attention to the very great effect of a factor which is frequently overlooked, viz., the gas content of the steel. The occasion which gave rise to their work was the behaviour of a nickel-chromium steel ingot cast from a melt of composition, carbon 0·15, chromium 0·6 to 0·9, nickel 3·5 per cent., made in an arc furnace. The melt was normal, the steel was killed with ferro-manganese and ferro-silicon, and 0·3 kilo. per ton of aluminium was added. The casting temperature was 1400 deg. by optical pyrometer, uncorrected. The first ingot showed the usual pipe, others had a slight tendency to rise. As the melt had been fully deoxidised, the assumption was made that the steel was near its saturation limit for hydrogen or nitrogen. A piped ingot was sectioned longitudinally and proved to be sound. A remarkable feature of its structure, however, was its large primary structure and tendency to columnar crystallisation. Grain boundary cracks were recognisable between the large primary crystals. The appearance was similar to that observed by H. Bennek, H. Schenck, and H. Müller[2] in ingots cast from steel to which hydrogen additions had been made.

In making further investigation of this matter, Hohage and Schäfer preferred not to use melts to which hydrogen had been added, as liberation of the gas during solidification causes turbulence which interferes with the normal development of the structure. It was desired to test the supposition that high gas content favoured columnar growth by observations on quietly solidifying melts. Accordingly a number of experimental 20 kilo. melts were made in a high-frequency furnace, some in the ordinary atmosphere and some *in vacuo*. Only melts of similar composition made from the same raw materials with the same conditions of melting were compared. The vacuum employed was about 1 mm. at the beginning of melting, rising to 10-15 mm. during melting, and was reduced to 1 mm. before the steel was cast. Water-cooled copper moulds were used, 450 mm. long, with internal diameter of 85 mm. at the top, 80 mm. at the bottom. After complete solidification the ingots were removed from the moulds and cooled in air. The types of steel investigated were of the following approximate percentage compositions :—

C.	Cr.	Ni.	Mo.
0·35	0·85	3·5	—
0·30	0·90	—	0·3
0·40	0·8	2·0	—
0·50	2·2	1·0	—
0·14	18·0	7·0	—

Both by fracture and by the etching of cross sections of the ingots it was established without exception for all these steels that all melts which had been melted in the open and cast in the open showed pronounced columnar crystals succeeded by coarse equiaxial structure in the centre of the ingot, while all melts which had been melted *in vacuo* and cast either *in vacuo* or in the open atmosphere showed no columnar crystals, but a fine equiaxial structure throughout. Casting in a thick-walled, water-cooled copper mould might have been expected to produce columnar crystals every time, but it never did so in the steels melted *in vacuo*. The two types were examined for high melting point non-metallic inclusions which might possibly have acted as nuclei, but they were similar in this respect. In any case, a low inclusion content, such as might have been expected in the vacuum-melted steel, would tend to favour continuance of columnar growth rather than its suppression. Deoxidation by carbon can proceed to completion *in vacuo* as the carbon monoxide produced according to the equation $FeO + C \rightarrow Fe + CO$ is removed as soon as it is formed. A number of open melts were deoxidised by ferro-silicon and an excess of the strongest deoxidising agents was added, *e.g.*, up to 0·6 gramme Al per kilogramme, and up to 0·3 gramme Ti per kilogramme, but columnar growth in the ingot was neither removed nor diminished. The artificial production of nuclei in the form of aluminium nitride and titanium nitride did not cause disappearance of the columnar structure. High casting temperature produced marked columnar crystallisation and low casting temperature a smaller

development of columnar crystals. This can hardly be attributed to different degrees of deoxidation, but may well be due to different gas contents. In this connection reference is also made to the greater susceptibility of hot casts to grain-boundary cracks and to flakes. The authors hold that the vacuum treatment is determinative in the avoidance of columnar crystallisation. Melts with a high gas solubility, such as high alloy steels of low carbon content, must be held for a correspondingly longer time under a high vacuum. They conclude that to promote columnar growth in steel, two conditions are essential—a suitable temperature gradient and a certain gas content. One of these alone is not sufficient. Columnar growth is not possible without a suitable temperature gradient; on the other hand, gas-free melts solidifying even in a steep temperature gradient show no columnar crystallisation.

The authors picture the process as occurring somewhat on the following lines:—As the columnar crystals grow the gases present (mainly hydrogen) segregate towards the centre of the ingot, the gas enrichment exercising a depressing effect on the liquidus temperature, until the layer adjacent to the growing crystals is saturated. When the saturation point is just exceeded a local turbulence prevents the further growth of the columnar crystals. The remainder of the ingot then solidifies into equiaxial crystals. The saturation limit for gas is only reached in the layers which mark the boundary of the columnar crystals. There is no rising of the steel; it is only when the remaining liquid core of the ingot is saturated that any visible disturbance occurs.

According to these ideas the kind of crystallisation is determined in a large degree by the gas content of the steel. Steel cast at a low temperature shows little columnar crystallisation because its gas content is low. Steel cast at a high temperature shows a considerable depth of columnar crystal because its gas content is high, and, moreover, in line with the hydrogen theory of flake formation, it is approximately at the boundary of the columnar region that flakes are most prevalent in steels which contain this defect. Similarly, steels cast in moulds without any dressing show less columnar crystallisation than those cast in moulds dressed with materials capable of giving off gases, especially cracked hydrocarbon gases. The influence of the composition of the steel is also regarded as indirectly determined by its capacity for dissolving hydrogen.

An analogy is drawn between the process of solidification of rimming steels and killed steels. In both cases columnar crystal growth is accompanied by enrichment of the boundary layer between solid and liquid, in rimming steels by enrichment in hydrogen and FeO, in killed steels in hydrogen only. In killed steels the hydrogen remains in solution, but in rimming steels the reactions $FeO+C \rightarrow Fe+CO$ proceeds and the carbon monoxide, being far less soluble than hydrogen, is liberated and entrapped in blow-holes. Just as the gases in a rimming steel leave their distinctive mark on the macrostructure of the ingot, so also, according to the authors, does the hydrogen content of a killed steel exercise a no less characteristic effect on the process of solidification. It may be considered that the authors overemphasise the part played by hydrogen in relation to that of other variations of composition brought about by microsegregation during solidification (which, however, they would claim to be effective in so far as they influence hydrogen solubility), but the important experimental fact is established that vacuum melts of the alloy steels employed solidified, even in water-cooled copper moulds, to fine equiaxial grains, while the same steels melted under similar conditions except that they were exposed to the atmosphere during melting showed pronounced columnar crystal growth. This fact is one which must be taken account of in the planning and interpretation of all experimental work on the occurrence of columnar structure in steel ingots.

References

[1] *Archiv für das Eisenhüttenwesen*, September, 1939, 1939–40, 13, 123.

[2] *Stahl und Eisen*, 1935, 55, 321.

Intercrystalline Corrosion of Chromium-Manganese Steels

The high cost of the 18 : 8 chromium-nickel austenitic corrosion-resisting steels has led to proposals for a reduction in amount or the substitution of the alloy elements employed. No element can, so far as it is at present known, replace chromium in conferring corrosion resistance, and since the ferritic chromium steels are liable to coarse grain size and embrittlement, it is necessary, if the best properties are to be obtained, for the steel to be in the austenitic condition. The austenitic structure may be induced by manganese as well as by nickel, but an 18 : 8 chromium-manganese steel contains both austenite and ferrite, and its corrosion resistance is less than that of 18 per cent. chromium steel and much inferior to 18 : 8 chromium-nickel. By reducing the chromium and increasing the manganese it is possible to attain the austenitic state and with it mechanical properties very similar to those of the 18 : 8 chromium-nickel steel.[1] The diminution in chromium content reduces the general resistance to corrosion, and should therefore be as small as possible. A 15 : 9 chromium-manganese steel is austenitic after quenching, but the austenite is not stable, and the steel becomes hard and magnetic on being cold worked, owing to the formation of martensite. To obtain a more stable austenite it is, as the structural diagram indicates, useless to increase the manganese content. Two solutions are possible:—

(1) To reduce the chromium still further and use 8 : 18 chromium-manganese steel. This steel, however, does not possess sufficient resistance to corrosion to justify using the large amount of alloy elements present.

(2) To stabilise the austenite by the addition of a small amount of nickel, *e.g.*, by using 15 : 12 chromium-manganese with 1 or 2 per cent. of nickel or 15 : 9 chromium-manganese with 2 or 4 per cent. of nickel.

The three types of chromium-manganese steels, 18 : 8, 15 : 12 containing nickel, and 15 : 9 containing nickel have been the subject of investigation by A. de Sy[2] to determine their susceptibility to intercrystalline cracking and to compare this phenomenon in chromium-manganese steels with the corresponding effect which is known in the plain 18 : 8 chromium-nickel steels as weld decay.

In the chromium-nickel austenitic steels the γ solid solution is supersaturated with carbon at ordinary temperatures. On reheating to a sufficiently high temperature, the excess carbon is deposited in the grain boundaries and on slip planes in the form of chromium carbide, Cr_4C, having taken up the chromium required for the formation of this com-

pound from the adjacent solution. Below 800 deg. Cent. this local impoverishment of the solution cannot be compensated by diffusion and the local resistance to corrosion is therefore seriously diminished. The maximum effect is produced by reheating for a short time at about 700 deg. Cent. The solid solution at the grain boundaries becomes impoverished in chromium and liable to attack both by electrolytic action and by simple solution by a corrosive agent which may have no effect on the true 18 : 8 solid solution. A suitable reagent is the well-known solution containing 10 per cent. sulphuric acid +10 per cent. copper sulphate, which at boiling point will attack the low-chromium grain boundaries and disintegrate the steel.

The austenitic chromium-manganese steels after having been reheated to about 700 deg. Cent. are liable to the same kind of intercrystalline attack. Microscopical examination of sections electrolytically etched in 10 to 15 per cent. chromic acid solution was made by de Sy. The grain boundaries were first revealed as a discontinuous network, the most impoverished areas being attacked first. With more prolonged etching the boundaries appeared to become wider and the network more complete as the attack proceeded to the regions in which the chromium content had not been so greatly reduced.

Loss of weight tests were carried out, with results shown in Table I. The attack was intercrystalline.

TABLE I.—*Loss of Weight in Grammes per Square Metre per Hour of Chromium-Manganese Steels Immersed for Twenty-four Hours in Boiling 10 per cent. H_2SO_4 +10 per cent. $CuSO_4$ Solution*

Steel. No.	Composition, per cent.					Reheating treatment.	Loss of weight, g/m²/hr.
	C.	Cr.	Mn.	Ni.	Cu.		
A	0·15	14·80	10·00	2·15	—	—	1·2
						1 hr. at 725 deg.	44·3
B	0·12	16·10	10·12	2·12	—	—	1·2
						1 hr. at 725 deg.	20·8
C	0·08	18·35	9·05	—	0·85	—	0
						1 hr. at 725 deg.	0
						1 hr. at 650 deg.	4·1
						1 hr. at 600 deg.	39·6
D	0·10	14·07	8·82	1·55	0·33	1 hr. at 725 deg.	79
E	0·12	15·24	11·82	2·10	0·50	1 hr. at 725 deg.	87

All the austenitic steels (*i.e.*, all the steels except C, which was half ferritic) lost their metallic ring. Steel C did not do so until after 200 hours, as the network of precipitated carbides was interrupted by the ferrite and the cracks took longer to penetrate through the specimen. In this mixed structure of ferrite and austenite the maximum attack occurred after reheating at 600–650 deg. Cent. for one hour. The extent of intercrystalline cracking can quickly be assessed by bend tests on the reheated steel after immersion for a short time in the boiling H_2SO_4 +$CuSO_4$ solution. Steels A and B showed greatest susceptibility after being reheated at 650 deg. or 750 deg. Cent., and immersed three hours in the solution. Steel C required a longer period of attack, and after 40 hours' immersion showed a greater effect in the specimen which had been reheated at 650 deg. Cent. than in those reheated at 550 deg. or 750 deg. Cent. In cold-worked material of each composition the effect was considerably delayed. A flat specimen, stamped out cold from a plate of steel D, reheated at 725 deg. Cent., and treated in the H_2SO_4+$CuSO_4$ solution for 24 hours, remained intact all round the edges which had been sheared, while the central part of the specimen which had received no cold work could be disintegrated by the finger. The depth of penetration of the attack was also followed in the different steels by microsections cut normally to the surface.

The idea that preferential attack takes place on solid solution impoverished in chromium was supported by the results of analysis of the total metal dissolved by the corrosive reagent. Thus the percentage of chromium in the total metal going into solution gradually rose as the time of immersion increased (Table II). Carbide did not appear to

TABLE II.—*Progress of Attack on Specimens (weight, 69 grammes, thickness 2 mm.) of Chromium-Manganese Steel E (Chromium, 15·24 per cent.) Immersed in Boiling 10 per cent. H_2SO_4+10 per cent. $CuSO_4$ Solution*

Period of immersion.	Duration, hours.	Loss of weight, per cent. per hour.	Chromium per cent. in dissolved metal.
0– 1 hour ...	1	0·068	7·48
1– 3 hours ...	2	0·058	8·08
0– 6 ,, ...	6	0·068	9·07
6–12 ,, ...	6	0·100	11·42
12–24 ,, ...	12	0·200	14·50
0– 1 hour ...	1	0·070	7·00
0– 3 hours ...	3	0·067	7·91
0– 6 ,, ...	6	0·080	9·53
6–12 ,, ...	6	0·087	12·02
0–12 ,, ...	12	0·095	10·51
0–24 ,, ...	24	0·100	13·04

become detached from the specimen until after the twelfth hour, at which time the attack on the grain boundaries of this particular steel had been completed and general disintegration had begun. The manganese content found in the total metal dissolved was always higher than that in the steel; the nickel content, uncertain on account of the small amount, probably about the same. The actual values will be affected by temperature and duration of the reheating treatment, and also probably by the form of specimen (superficial area in relation to volume).

The austenitic chromium-manganese steels are subject, like the chromium-nickel steels, to intercrystalline corrosion and the work of de Sy indicates a close parallel in the conditions which produce susceptibility to the defect. The susceptibility is present also in the 18 : 8 chromium-manganese steel, which contains austenite and ferrite, though the effect is shown more slowly. The paper does not go on to indicate how the danger may be avoided, but in view of the analogy established with the chromium-nickel steels, it may be assumed that a minimum possible carbon content, together with the addition of carbide precipitating elements, would eliminate the defect.

REFERENCES

[1] For references to the constitution and properties of chromium-manganese steels, see THE METALLURGIST, 1937–38, 11, pages 11, 123, and 158; and 1939 (October), page 70.
[2] *Revue de Métallurgie*, 1939, 36, 389.

The Time Factor in the Tensile Testing of Austenitic Steels

THE influence of speed of testing on the results of tensile tests has been the subject of experiment by many investigators. In general it has been found that, not only the elastic limit and yield point, but

also the ultimate stress, the elongation, and the reduction of area are all increased in different proportions by the performance of the tensile test at high speed. For example, F. Fettweis[1] obtained the following results in a basic steel of medium hardness: As the duration of the test was reduced from twenty-four minutes to nine seconds the yield point was raised 3·5 per cent., the ultimate stress 7·2 per cent., the elongation 17·6 per cent., and the reduction of area 18·7 per cent. Results obtained by other investigators were compared by Fettweis, who drew the conclusion that with increase in the rate of loading both yield stress and ultimate tensile strength were always raised. The effect of speed on elongation and reduction of area is not so regular, but on the whole it is generally held that elongation and reduction of area do not suffer by very rapid straining.[2]

While the above generalisation may be true for the yield stress, it is probable that the relation between rate of strain and ultimate tensile strength is by no means so simple. Throughout the period which elapses between the moment of exceeding the yield point and of fracture of the test piece, the material is being progressively overstrained and at the same time is undergoing strain ageing. If a definite amount of overstrain is applied to a specimen, its properties will for a time undergo spontaneous change, in the direction of restoration of elasticity and a further hardening effect, as a result of ageing. Similarly, with the same period of ageing, different amounts of overstrain produce different effects. The relation between the yield point and the maximum load reached before necking (with a possible consequent effect on elongation and reduction of area) is therefore dependent primarily on the strain-age-hardening properties of the material at the temperature of the test. At temperatures at which strain-ageing will take place with sufficient rapidity, hardening occurs throughout the whole cross section of the test piece and the maximum load attained is thereby increased. For example, in a medium carbon or alloy steel, although the yield point is always raised by increased rate of strain and, at room temperatures, the maximum load in the rapid test is higher than in the slow test, it is less in the rapid test than in the slow test at 200 deg. Cent. The elevated temperature accelerates the ageing process and the specimen becomes strain-age-hardened during the period between the yield and the break. With Armco iron ageing is more rapid. The slow test gives a higher maximum load than the quick test at 100 deg. Cent., and even at room temperature the extra speed does not result in a higher maximum load than is produced by the strain ageing which accompanies the slow test. With Armco iron it is necessary to carry out tests well below 0 deg. Cent. in order to be sure of obtaining increased tensile strength with increased rate of strain.

This dependence of the time factor in tensile testing on the strain-age-hardening properties of the material makes it specially interesting to observe the behaviour of certain austenitic steels which are known to be much more susceptible to strain hardening than are the carbon steels.

TABLE I.—*Chemical Composition of the Steels*

Mark.	Composition, per cent.					Treatment.	Structure.
	C.	Si.	Mn.	Cr.	Ni.		
A1	0·07	0·54	0·38	18·50	8·50	1050 deg. water	Austenite, fine grain
A2	,,	,,	,,	,,	,,	1250 deg. water	Austenite, coarse grain
B	0·07	0·09	0·24	—	27·58	1150 deg. water	Austenite
C	0·28	0·33	0·59	12·55	8·35	1150 deg. water	Austenite and carbide
D	0·16	0·40	0·33	10·94	31·96	1150 deg. water	Austenite
E	0·54	0·99	0·79	21·33	19·58	1100 deg. water	Austenite and carbide
F	1·08	0·33	13·35	—	—	1100 deg. water	Austenite
G*	0·46	0·91	0·49	26·62	0·36	860 deg. air	Ferrite and carbide

* G also contained 0·2 per cent. molybdenum.

TABLE II.—*Tensile Properties in Relation to Rate of Straining*

Mark.	Rate of straining, mm. per min.	Duration of test.	Yield point, Tons per sq. in.	Maximum load, Tons per sq. in.	Elongation per cent., $l = 10\,d$.	Reduction of area, per cent.
A2	0·5	185 min.	12·0	48·9	74	64
	2	40 min.	12·7	42·5	65	70
	5	15 min.	13·3	40·0	61	68
	25	170 sec.	14·0	38·7	58	70
	50	85 sec.	16·5	38·1	58	71
	90	50 sec.	19·0 (*a*)	38·1	57	71
E	0·5	70 min.	26·0	48·9	29	48
	2	17½ min.	26·7	48·9	28	47
	5	7 min.	28·6	48·2	28	47
	25	80 sec.	30·5	49·6	27	44
	90	25 sec.	32·4 (*b*)	48·9	30	49
F	0·5	105 min.	22·2	60·3	42	28
	2	30 min.	22·2	61·0	45	30
	5	12 min.	22·2	62·8	46	30
	25	2¾ min.	24·8	64·1	52	34
	50	85 sec.	24·8	64·1	57	35
	90	50 sec.	29·2 (*c*)	64·1	52	34

(*a*), (*b*), (*c*), upper yields; lower yields, (*a*) 17·8, (*b*) 30·5, (*c*) 27·3.

J. Musatti and E. Hugony[3] have made an extensive research on this subject, selecting for investigation several different austenitic steels and for comparison a ferritic chromium steel and an Armco iron. The composition of these steels and the heat treatment applied to them is given in Table I. The specimens employed were 10 mm. in diameter with an acting length of 120 mm. The rate of straining was varied between 0·5 mm. and 90 mm. per minute, the duration of the tests being from 180 min. to 40 sec. Steels B and C lie on the boundary of the austenite-martensite region of the structural diagram of the chromium-nickel steels proposed by E. Maurer. The austenite of which they consist will therefore readily undergo a change to martensite. Steel A is not far removed from the same region. Steels D and E, however, are well within the region of stable austenite. Their behaviour on cold working will therefore be different. Steels A, B, and C, after deformation, will form martensite with consequent marked hardening, whereas the cold-worked steels D and E might be expected to be relatively free from this effect. Typical examples of the tensile properties are shown in Table II, each value representing the mean of two experiments, the results of which showed good agreement. For steel A, water quenched from 1250 deg. Cent., and therefore possessing a coarse grain, the yield point increased continuously with increased rate of straining, while the tensile strength fell rapidly at first and then approached a constant value as the rate of straining was increased. It seems quite possible that at a still more rapid rate it would rise again. The elongation showed a marked fall until the rate of straining reached 25 mm. per minute, and was then constant for faster rates. The reduction of area was fairly constant throughout. Structurally it was found that the martensite formation was greatest with the slow rate of strain, and the high value of the tensile strength with a rate of strain of 0·5 mm. per minute is readily explained by the structural changes undergone (Figs. 1 and 2). Steel A in the fine grain condition and steels B and C behaved similarly. Steel E (consisting of stable chromium-nickel austenite) showed no similar effect of slow straining on the maximum load, though the yield point rose with increased rate of straining, in the usual way. Elongation and reduction of area were approximately constant. In the 13·3 per cent. manganese steel F, which, according to Guillet's diagram, lies well within the austenitic region, the yield point rises with increased rate of straining and so also does the tensile strength. Elongation and reduction of area also show a slight rise with increased speed of testing. The slow tests give the lower values throughout.

These tests are of interest in showing the behaviour of unstable and stable nickel-chromium austenite and of stable manganese austenite. It will be understood, in the light of the opening comments of this article, that the results should be regarded as applicable to the conditions of test only, in particular to behaviour at room temperatures only. To investigate the effect of the time factor in the tensile testing of these steels over a wide range of temperatures would probably yield results of great interest. Any factor which might affect the rate of strain hardening of the steel would be likely to modify the effect of different rates of testing.

FIG. 1—*After Quenching from* 1250 *Deg. Cent. in Water*

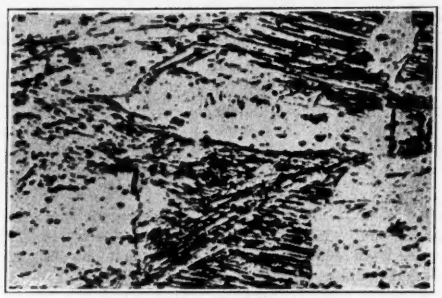


FIG. 2—*After Very Slow Straining*

Structure of Steel with C 0·07, *Cr* 18·5, *and Ni* 8·5 *per Cent. Magnification* 200 *Diameters*

REFERENCES

[1] *Archiv für das Eisenhüttenwesen*, 1932–33, 6, 149; see also THE METALLURGIST, 1932, 8, 179.

[2] R. J. Lean and H. Quinney, "Proc.," Inst. Mech. Eng., 1937, 135, 467.

[3] *Metallurgia Ital.*, 1938, 30, 545; see also *Stahl und Eisen*, 1939, 59, 821, and *Revue de Métallurgie*, 1939, 36, 273.

Sources of Failure in Hot-Pressing of Zinc Alloys

LITTLE has been published on the hot-pressing process as applied to the alloys of zinc, which is at present only in a very early stage of development. An informative and well-illustrated article on the various troubles met with in the process should therefore be specially valuable. A German paper by W. Wolf[1] can be particularly recommended to the attention of those engaged in hot-working zinc alloys. It shows that experience must be gained in these new alloys and that, to achieve successful results, their characteristics must be studied.

The subject is of further interest as zinc has the same kind of space-lattice structure as magnesium. Magnesium with an hexagonal structure requires a special forging technique of its own, in view of the marked directional properties of the metal. Aluminium, which is only slightly heavier than magnesium, is a light metal with a cubical structure, and does not show these special forging characteristics.

An important factor in the working of zinc alloys is careful adherence to predetermined temperatures of operation. It is also found that deformation of the

[1] *Zeitschrift für Metallkunde*, 1939, Vol. 31, 69–72.

metal proceeds better when carried out at slower speeds. In producing hot pressings, forging must be done, as far as possible in a direction perpendicular to the length of the fibres of the metal in the piece of bar or rod used. The low temperatures of deformation (about 300 deg. Cent.) necessitate attention to precautions, unnecessary in the case of brass.

A knowledge of the possible troubles which may be encountered is an assistance to those attempting to produce satisfactory pressings. During the last few years the hot-pressing of zinc alloys has been increasing, since the properties obtained are similar to those of brass and consequently the zinc alloys may be used as a substitute for that metal. The pressing and drawing operations, as well as forging in a die, are essentially the same as for other metals. However, characteristics natural to zinc alloys require attention, as is shown by the following troubles met with in their hot working.

As with other metals, rods, profiles, and tube made by extrusion require the use of ingots free from defects. The chief defects encountered in the casting of ingots for pressing are pipe, fine threadlike cavities, and contraction cavities. After-feeding of the ingot should have special attention, as zinc alloys are specially prone to suck up the molten metal on account of their high contraction on solidification. The presence of hollow spaces and cavities in the ingot is more dangerous than with brass, since they do not weld up on subsequent working in view of the lower temperature used for pressing (250 deg. to 300 deg. Cent.), and hence defects in the extruded material may be anticipated. Apart from inner defects, the extruded bar itself may exhibit surface troubles. One of the most common of these is longitudinal splitting of the bar caused by extruding too quickly. Previously, zinc alloys were extruded at a speed of 2 m. per minute, but nowadays the tendency is to extrude at speeds of 4 m. to 5 m. per minute, as a result of accumulated experience in the alteration of dies and suitable press operation. There is, however, a danger of cracks being present, which cannot be seen during the extrusion process and which cannot be entirely eliminated. They may show up when the bars are turned in a lathe. On subsequent drawing these cracks can lead to fracture of the material. The trouble has mostly been got rid of as a result of experience.

Surface blistering of the extruded bar may be due to oil, from the press parts, being forced (during extrusion) below the surface of the bar, after which it becomes vaporised and gives a swelling. In this case the blisters are mostly spread uniformly over the whole exterior of the bar. Blister trouble may be also traceable to the presence of blow-holes due to gas, mostly near the exterior surface of the ingot. This trouble is almost entirely removed by turning off the surface of the billet ("scalping"). Finally, with certain types of alloy, an air cushion may be formed inside the press and near the die, resulting in surface blistering of the extruded product. Fortunately, this trouble is soon detected and the defective parts discarded. Again, impurities such as oxides in the metal only make their presence felt in the last stages of fabrication, when the material is being surface-treated in the electrolytic bath. This can be avoided by cleanliness of working when the alloys are made.

When zinc alloys are extruded too quickly and when too much of the billet is pressed through the die there is a tendency for the middle of the billet to be forced through the die sooner than the outer part, with the result that the last piece to leave the die is hollow in section. Oxide films in the ingot will result in a tube being cracked after extrusion; the trouble was traced conclusively to "notch effect" caused by an inner accumulation of oxide particles. A micro-section showed that the crack followed the edge of the oxide skin. It appears that the oxide is not serious where the extrusions are of a massive nature.

In hot stamping the micro-structure of the zinc alloy after leaving the extrusion press is of considerable importance. Thus a zinc alloy with 4 per cent. aluminium, 0·5 per cent. copper, and 0·02 per cent. magnesium before extrusion consists of the zinc α solid solution surrounded by α and β eutectic. As would be expected, the size of the primary crystals (solid solution) depends on the casting conditions. If a micro-section be taken after extrusion in a direction perpendicular to the length of the bar and another

TABLE I.—*Mechanical Properties of Die-stamped Tensile Bars of Zinc Alloys*

Material...	ZL 3.		ZL 7.	
Analysis...	4% Al, 0·5% Cu, 0·02% Mg		4% Cu, 0·2% Al	
Testing conditions	Pressed parallel to the direction of the fibres in the blank.	Pressed perpendicular to the direction of the fibres in the blank.	Pressed parallel to the direction of the fibres in the blank.	Pressed perpendicular to the direction of the fibres in the blank.
Tensile strength, tons per square inch	20·3 to 22·2	24·1 to 25·4	19·0 to 20·3	21·0 to 22·9
Elongation, per cent. (l=10d)	4 to 6	8 to 11	10 to 13	15 to 20
Elongation, per cent. (l=5d)	7 to 10	13 to 16	8 to 12	18 to 25
Reduction of area, per cent.	10 to 20	20 to 40	10 to 20	25 to 50
Brinell hardness number (62·5 kilos./2·5 mm./30 sec.)	90 to 100	95 to 110	80 to 90	90 to 100

micro-section along the length of the bar, it will be found that the primary crystals are broken up into uniformly distributed small irregular areas, whilst in the longitudinal micro-section they appear as rough strings extending right across the section.

The machines used for the hot-pressing of zinc alloys are the same as those used for brass and light metals. Those machines which operate at a slower rate are, in general, better for working zinc alloys. In the manufacture of high-duty forgings it is advantageous when hot-pressing metals having a cubical space-lattice structure to place the blank, as cut from the extruded bar, in the die and work the material in the direction of the fibres (*i.e.*, in the longitudinal direction). It has been shown that a zinc alloy, containing either 4 per cent. of aluminium or 4 per cent. of copper, should not be treated in this way. A large number of experiments have shown that it is definitely better to press the alloy in a direction perpendicular to the fibre direction; that is, the piece of extruded bar is laid flat in the die. Higher mechanical properties for the hot-pressing are obtained in this way, and, on the other hand, the formation of cracks is avoided when the operation is so carried out as to give an even flow of the material in the die. Tensile bars forged perpendicularly to the fibre direction were found to have a tensile strength 15 to 20 per cent. higher and nearly double the elongation compared with tensile bars hammered parallel to the fibre direction (Table I). Moreover, there is a further

advantage that the flow of the metal in the die is more favourable.

When forging parallel to the longitudinal direction there is the possible danger of the formation of cracks, due to the different layers being pushed against each other during the process. Fig. 1 shows two hot pressings; the one on the left-hand side was forged parallel to the direction of the fibres, and the one on the right-hand side was forged perpendicularly to the fibres. The former shows several cracks, and, moreover, the metal has not filled the die completely, whereas that on the right is a perfect pressing. Micrographic examination showed that the flow of the metal was unsteady in the former case, but steady where forging occurred perpendicular to the direction of the fibres in the blank. It is not

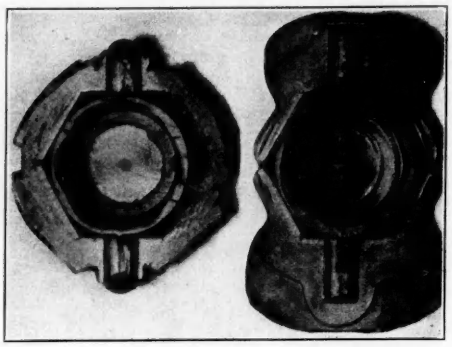


FIG. 1

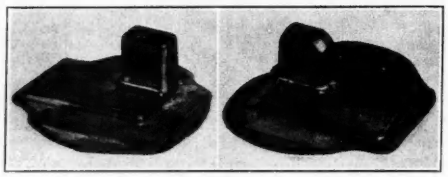


FIG. 2

FIGS. 1 AND 2—*Hot Pressings of Zinc Alloys; Pressed (Left) Parallel and (Right) Perpendicular to the Direction of the Fibres in the Blank*

always possible to lay the blank in this direction in the die, especially when producing hot pressings of the type shown in Fig. 2. The pressing on the right-hand side has not completely filled the die; in this case pressing was carried out perpendicular to the fibres of the metal. When the blank was laid longitudinally in the die, a satisfactory pressing was produced as shown on the left-hand side. With pressings of unsymmetrical shape, there is always the danger that distinct layers may be produced which flow over each other, and this may result in cracks being formed by one layer flowing more quickly over another layer. The defect may be remedied by carrying out the pressing operation in several stages. This is very necessary where pressings are to be subject to high service stresses because if the fibres in the piece do not lie in the right direction, a crack may appear owing to the inherent weakness of the metal.

Hot-pressing of zinc alloys needs careful attention. It is necessary to ensure that the blank is uniformly heated in the furnace before it is inserted in the dies. A difference of some 10 deg. Cent. between one part of the blank and another part is sufficient to cause uneven flow in the dies. A furnace provided with air circulation to give uniform temperature distribution is most desirable. It is also necessary to ensure that no swarf adheres to the blank after it has been sawn off, or it will result in a surface defect on pressing. In hot-pressing of brass, any swarf adhering to the blank would become welded on owing to the high temperature employed, but no welding occurs with zinc alloys as the temperature is not high enough.

Chromium-Molybdenum Steels

CHROMIUM-MOLYBDENUM steels are used for a wide variety of purposes in automobile, aircraft, and general engineering construction. They fall into several well-recognised groups, of which the one with the

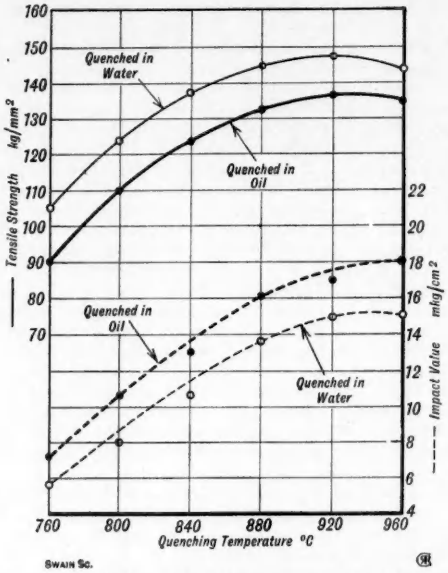


FIG. 1—*Influence of Quenching Temperature on Tensile Strength and Impact Value of Steel No. 1 (C 0·14, Mn 0·81, Cr 1·08, and Mo 0·28 per cent.). The specimens were tempered at 150 deg. Cent. for two hours after quenching*

widest application is that containing steels with about 1 per cent. of chromium and various carbon contents. In American and German specifications these steels are again divided according to their carbon content, *e.g.*:—

S.A.E. No.	C, per cent.	Mn, per cent.	Cr, per cent.	Mo, per cent.
4130	0·25 to 0·35	0·50 to 0·80	0·50 to 0·80	0·15 to 0·25
X 4130	0·25 to 0·35	0·40 to 0·60	0·80 to 1·10	Ditto
4135	0·30 to 0·40	0·60 to 0·90	Ditto	Ditto
4140	0·35 to 0·45	Ditto	Ditto	Ditto
4150	0·45 to 0·55	Ditto	Ditto	Ditto

Though the lower carbon steels (4130 and X 4130) are sometimes used as normalised, air-hardening is accentuated by higher carbon content, and the others are used only in the hardened and tempered condition. Chromium-molybdenum steels are often considered unsuitable for carburising on account of their depth-hardening properties, which may result in insufficient ductility in the core, but steels of lower carbon content than the above (*i.e.*, below 0·20 per cent. carbon) are used as case-hardening steels and are provided for as ECMo 80 and ECMo 100 in the German provisional specification DIN 1663. A very full account of the American steels is given in J. L. Gregg's "Alloys of Iron and Molybdenum."

The primary object of adding molybdenum to chromium steels is to decrease the effect of mass, and it is with this aspect of the subject that H. Voss and F. Krämer[1] are mainly concerned in their paper on "Attainable Mechanical Properties in Chromium-Molybdenum Steels." In this they report on the influence of mass, in bars of 5 mm. to 120 mm. diameter, on the properties of various experimental and commercial chromium - molybdenum steels quenched in oil and in water. In the first place, they studied a series of six commercial basic electric furnace steels with increasing carbon content (Nos. 1 to 6) and two series of experimental steels (Nos. 7 to 10 and Nos. 11 to 15).

Nos.	C, per cent.	Mn, per cent.	Cr, per cent.	Mo, per cent.
1 to 6	0·14 to 0·50	0·7	1·05	0·25
7 to 10	0·20	0·32 to 2·14	1·05	0·25
11 to 15	0·20	0·7	0·56 to 3·21	0·25

As a preliminary the most suitable hardening temperature was determined on 5 mm. diameter and 10 mm. square specimens cut from the outside of an 80 mm. diameter bar and tested separately. This temperature was found to fall from

Deg. Cent.			Per cent.
920 to 765	as the carbon	rose from	0·14 to 0·50
900 to 810	,, manganese	,,	0·32 to 2·14
870 to 850	,, chromium	,,	0·56 to 3·21

The curves for steel No. 1 (carbon 0·14 per cent.) are given in Fig. 1. The quenching was in each case followed by tempering for two hours at 150 deg. Cent. With rise in quenching temperature there was an increase both in the tensile strength and in the impact figure of this low carbon steel. Both properties reached a maximum value at about the same quenching temperature, and the impact values were surprisingly high. They were obtained on a 10 mm. by 10 mm. test piece, having a notch 2 mm. in diameter and 3 mm. deep. Ferrite was present in the steels quenched from the lower temperatures, but those quenched from 880 deg. Cent. and above were uniformly martensitic in structure. The highest values of tensile strength attained on oil hardening and tempering at 150 deg. Cent. are shown in Fig. 2. The influence of carbon content is considerable, while that of manganese and of chromium in the 0·2 per cent. carbon steel is small. The effect of manganese and of chromium on the hardness attainable in higher carbon steels was not studied. From these fifteen steels, specimens 5 to 120 mm. in diameter and of length (2 d+20) mm., were oil-hardened from 830 deg. to 840 deg. Cent., and tempered at 150 deg. Cent. for two hours. A section, 20 mm. thick, was cut from the middle and hardness tests made at centre and circumference to study uniformity of hardening. The greatest difference between the 5 mm. and the 120 mm. diameter specimens was shown by the medium carbon steels of the first series, and the least by the high chromium steels of series 3. With increasing carbon the difference in hardness between centre and outside increased; with increasing manganese and chromium it diminished.

Data were obtained from eighty-nine commercial case-hardening steels, sixty basic open-hearth and twenty-nine basic electric melts. It was found that with approximately the same composition (carbon 0·13 to 0·17, manganese 0·8 to 1·1, chromium 1·0 to 1·3, and molybdenum 0·2 to 0·3) a wide variation in tensile strength was obtained, the variation amounting in the 5 mm. diameter bars to 110 kilos. to 150 kilos. per square millimetre, and in the 30 mm. diameter bars to 80 kilos. to 115 kilos.

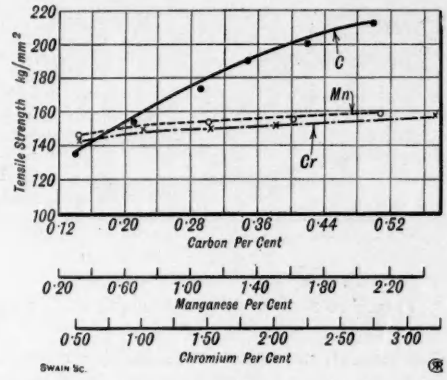


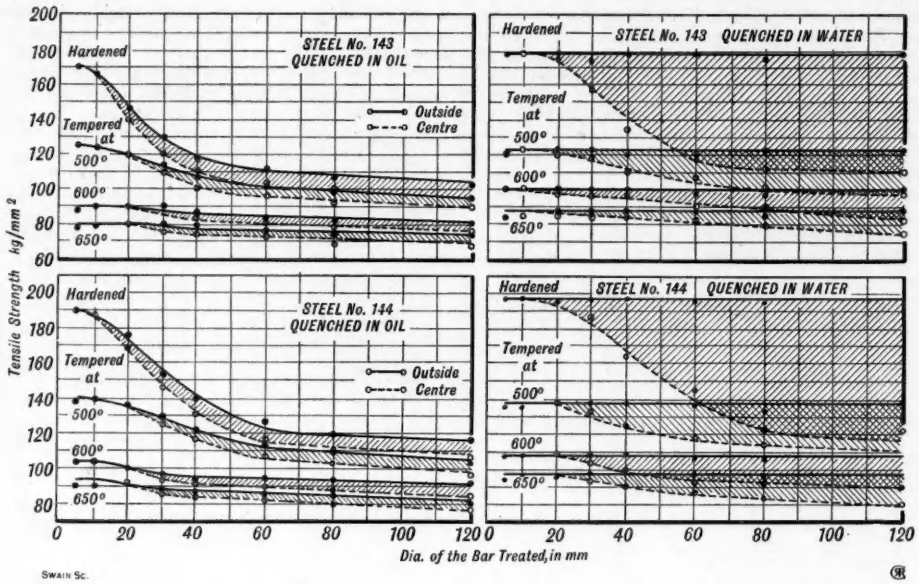
FIG. 2—*Influence of Carbon, Manganese, and Chromium on the Highest Tensile Strength Reached by the Oil-hardened Steels*

per square millimetre. In the higher carbon series (carbon 0·18 to 0·23 per cent.) the differences were even more marked. This wide variation in the properties of chromium-molybdenum steels of apparently the same composition has been recognised previously and some metallurgists have attributed it to variation in grain size. Specimens of twenty-eight melts having almost identical composition were therefore examined by the McQuaid-Ehn test for inherent grain size. This was found to vary between 3 and 8, but no clear relation between the tensile strength of the treated steel and grain size could be established. Neither was there any relation between tensile strength and mode of manufacture. It could only be concluded that the variation is brought about by some hitherto unrecognised factor.

In this case-hardening type of chromium-molybdenum steel containing about 0·2 per cent. of carbon, it is possible to obtain a good impact value in the hardened condition (oil-quenched from 830 deg. Cent. and tempered for two hours at 150 deg. Cent.). With increase in size from 10 mm. to 50 mm. in diameter the tensile strength in this condition falls

from about 88 to 64 tons per square inch. The impact value was 10–13 mkg. per square centimetre in the small section, falling to 6·5–8 mkg. in the 30 mm. diameter bar, and rising slightly, on account of fall of hardness, in the specimens of 50 mm. diameter. The microstructure, martensitic up to 20 mm. diameter, shows an increase in the amount of free ferrite as the size increases above that limit.

In the chromium-molybdenum steels of higher varieties, Nos. 143 and 144, water quenching gives complete hardening at the outside over a very large range of sizes, but there is a big difference in hardness between outside and centre in the large sizes. The lack of uniformity is not entirely obliterated, even with temperating temperatures as high as 650 deg. Cent., though the average hardness or tensile strength of the thick sections is much greater than that of the incompletely hardened specimens obtained



FIGS. 3 TO 6—*Influence of Cross Section on the Tensile Strength at Centre and Outside of Steels Nos.* 143 *and* 144

carbon content, the maximum attainable strength is greater, and there is a somewhat greater fall in tensile strength with increasing cross section than in the case-hardening steels. Chromium-molybdenum steels of this class, typical of those used for constructional purposes, were studied in a similar way to the case-hardening steels. Their compositions are given in Table I, and the results of tests are shown in Figs. 3

TABLE I.—*Composition of the Steels*

No.	C, per cent.	Mn, per cent.	Cr, per cent.	Mo, per cent.	Designation.
143	0·28	0·63	1·14	0·22	V C Mo 125
144	0·35	0·65	0·97	0·23	V C Mo 135
145	0·42	0·61	0·97	0·24	V C Mo 140
146	0·40	0·65	1·77	0·31	V C Mo 240
147	0·33	0·67	2·38	0·30	Flieg 1456

to 6. In each case the shaded region represents the range of tensile strength or hardness which characterises the section from outside to centre. The diagram indicates values of tensile strength, but these are derived from Brinell hardness numbers multiplied by 0·35. The properties are also shown after tempering at 500 deg., 600 deg., and 650 deg. Cent. The diagrams are useful in enabling an estimate to be formed of the possibility of attaining any required tensile strength in different sizes. In the low-carbon by oil quenching. In the other three steels water hardening cannot be used on account of the risk of cracking. The effect of the higher chromium content in Nos. 146 and 147 is shown by the higher tensile strength and greater uniformity across the section obtainable in the larger specimens.

All the steels are appreciably affected by the mass of the specimen and it is therefore desirable that all rough machining should be done before heat treatment. The authors give 60 mm. (2·36in.) as the limit of useful size for steel No. 143 after oil hardening and tempering to attain a tensile strength of 80 kilos. per square millimetre (50·8 tons per square inch), or for steel No. 144 to reach 100 kilos per square millimetre (63·5 tons per square inch).

For higher tensile limits or larger masses, it is necessary to increase the alloy content of the steel, as in steels Nos. 146 and 147. These are suitable for 65 tons per square inch tensile strength in 4in. and 5in. diameter bars respectively. The authors recorded experiments do not go beyond this stage. To obtain any further advance in the mechanical properties of large masses it would be necessary to increase both chromium and molybdenum contents, or to introduce some other alloy element, especially nickel, in order to bring about hardening throughout the section.

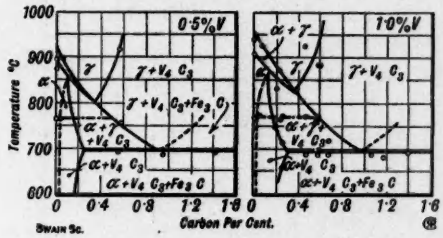
REFERENCE

[1] *Stahl und Eisen*, August 10th, 1939, 59, 913.

The Influence of Rate of Cooling on the Transformations in Vanadium Steels

THE important researches of F. Wever and A. Rose on the transformations occurring in carbon steels at different rates of cooling have been extended to cover the transformations and properties of vanadium steels and of chromium steels.

The main features of the results on carbon steels

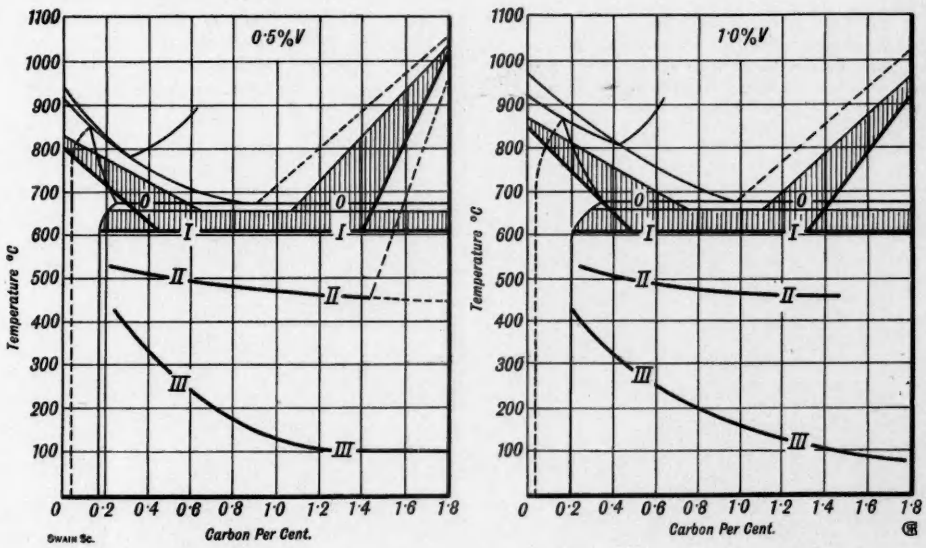


FIGS. 1 AND 2—*Sections Through the Iron-Carbon-Vanadium Diagram at 0·5 and 1·0 per cent. of Vanadium*

have already been described in THE METALLURGIST.[1] The relation between the temperature of the austenite transformation and the rate of cooling is summarised in undercooling diagrams. Those for carbon steels were reproduced in the previous article, and show, at certain critical rates of cooling, a sudden fall in the austenite transformation point from its equilibrium position at about 700 deg. to 650–600 deg. Cent., then to 500 deg. Cent., and finally to lower temperatures as the rate of cooling is increased.

In continuation of this work, Wever and Rose[2] have investigated two series of vanadium steels with vanadium 0·5 and 1·0 per cent. and carbon 0·3 to 1·7 per cent., both thermally and microscopically at increasing rates of cooling. The constitution of these two series of vanadium steels under equilibrium conditions is shown in Figs. 1 and 2. The undercooling diagrams are reproduced in Figs. 3 and 4, the upper unshaded lines corresponding to those shown in the equilibrium diagrams. It will be sufficient if the diagram for the 0·5 per cent. vanadium steels is referred to in detail; the interpretation of the other follows on parallel lines. On cooling at a rate of about 20 deg. per second, the pearlite charge is lowered from the equilibrium position (marked O) to about 600 deg. Cent., as shown in undercooling curve I (Fig. 3). The pearlite structure does not change in character except to become finer, but no V_4C_3 is deposited from the γ solution, ferrite precipitation is restricted to carbon below 0·5 per cent., and cementite to carbon above 1·4 per cent. The hardness increases rapidly with increased rate of cooling, while the structure remains pearlitic. There is no progressive fall in the temperature of the pearlite change (over 50 deg. or so) as in a carbon steel, but already with rates of cooling exceeding 15 deg. per second a new and independent transformation was observed, as shown in undercooling curve II. At this stage the separation of pre-eutectoid ferrite is suppressed, the temperature of the austenite transformation sinks to 525 deg. Cent. in steel of low carbon content and to 450 deg. Cent. in the 1·5 per cent. carbon steel. Cementite separation in the form of needles is evident in the high-carbon steels. The structure of the product of this intermediate stage is generally acicular, like slightly tempered martensite. At higher rates of cooling martensite appears at the temperatures shown in undercooling curve III. The temperature of the martensite change is the same as in carbon steel, and is independent of the cooling velocity beyond the minimum required to produce the undercooling. The structure of the martensite is similar to that of the martensite in carbon steels, but finer.

The hardness of the steels rises with increasing rate



FIGS. 3 AND 4—*The Undercooling Diagrams of Steels with 0·5 and 1·0 per cent. of Vanadium*

of cooling up to the end of the pearlitic stage at which it reaches a Vickers hardness of 400–500. In the intermediate stage the hardness falls, except in hypereutectoid steels, in which it rises continuously with increasing rate of cooling. The hardness rises in the martensitic product, up to 900–950 (Vickers) in the eutectoid steel, as the rate of cooling becomes more rapid, but at very high rates of cooling a diminution sets in as a result of the increased proportion of untransformed austenite.

The nature of the transformation of vanadium steels in the first stage (I) is similar to that of carbon steels, except for the precipitation of vanadium carbide. The rapid increase of hardness with increased rate of cooling in the first stage is due to the retention of V_4C_3 in the γ solution and the temper-hardening effect of its subsequent precipitation in the pearlite. No satisfactory explanation of the lowering of hardness which accompanies the intermediate stage is given. There is clearly an absence of the temper-hardening effect which occurs when the transformation occurs at the pearlite stage. Several steels were cooled at different rates, and then tempered at temperatures up to 750 deg. Cent. for half an hour. Those which had transformed at the pearlite stage (I) showed a marked increase in hardness at 400 deg. Cent., and the faster the initial cooling, the greater was the increase in hardness. In those initially cooled more slowly, some precipitation with coalescence had already occurred. Steels which transformed at the intermediate stage (II) showed no appreciable temper-hardening, but a slight rise in hardness after treatment at 400 deg. Cent. was followed by a sharp fall.

The martensite change (III) in vanadium steels was not different from the corresponding change in carbon steels. The critical cooling velocities required to retain a completely martensitic structure in the quenched material are shown in Fig. 5. They are very markedly lower than those of carbon steels.

A few experiments were made with very high vanadium contents, and illustrate the ineffective hardening of steels of these compositions. With 0·9 per cent. of carbon and 1·5 per cent. of vanadium a martensitic structure was obtained with a cooling rate of 230 deg. per second, but when the vanadium reaches 2 per cent. a rate of over 800 deg. per second is required. This is due to the formation of vanadium carbide, which is no longer taken into solution at the temperature of quenching, and has the effect of impoverishing the austenite of carbon. As a result, the hardness of a 0·9 per cent. carbon steel, quenched from 1100 deg. Cent., rises from 750 to 900, when 0·5 per cent. of vanadium is present, but as the vanadium content increases, falls slowly at first, and then very rapidly, to 200, when the vanadium reaches 2 per cent.

In Fig. 5 it will be observed that the critical cooling velocities of medium carbon steels containing 0·5 per cent. of vanadium were less than those of the 1 per cent. vanadium steels. This supports the widely accepted conclusion that vanadium is most effective in small amounts and that for ordinary constructional purposes no advantage can be expected in going beyond, or even as high as, 0·5 per cent.

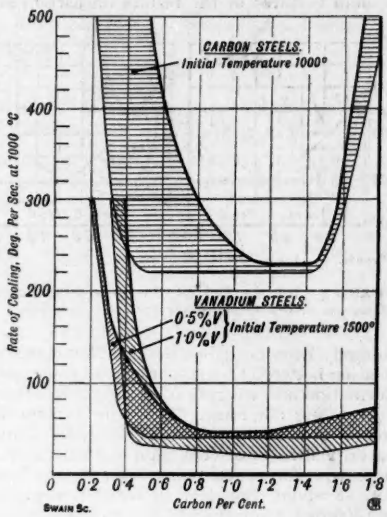


FIG. 5—*The Range of Critical Cooling Velocities in relation to Carbon Content for Carbon Steels and for Vanadium Steels*

REFERENCES

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